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TECHNICAL REPORT ARCCB-TR-88031

**LIQUID PHASE SINTERING OF CARBIDES  
USING A NICKEL-MOLYBDENUM ALLOY**

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**JULY 1988**

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**US ARMY ARMAMENT RESEARCH,  
DEVELOPMENT AND ENGINEERING CENTER  
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energy, and hardness using 20 percent binder with lesser amounts of molybdenum (6.25 or 12.5 weight percent) added to nickel compared to pure nickel. A refinement in the carbide microstructure and/or a reduction in porosity was seen for both the titanium and tungsten carbides when the alloy binder was used, compared to using the nickel alone. Curves depicting the above properties are shown for increasing amounts of molybdenum in nickel for each carbide examined. Loss of binder phase due to evaporation was experienced during heating in vacuum at sintering temperatures. In an effort to reduce porosity, identical specimens were processed by hot isostatic pressing (HIP) at 15 Ksi and at temperatures averaging 110° below the sintering temperature. The tungsten carbide and titanium carbide series containing 80 and 90 weight percent carbide phase, respectively, showed improved properties after HIP, while properties decreased for most other compositions.



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## INTRODUCTION

Carbides are extremely hard and brittle in pure form. Their melting points average from 3000°C to 4000°C with positive coefficients of expansion and high thermal conductivity. Carbides are usually combined with a binder phase such as cobalt, nickel, and in some cases ferrous alloys. Liquid phase sintering above the melting point of the binder results in densification needed to achieve useful properties. The binder usually has little solubility in the hard phase, or carbide phase, but may be appreciable in certain systems. The use of nickel instead of cobalt as a binder or alloy base, combined with carbides, can be an important factor in conserving strategic metals such as cobalt.

The microstructure resulting from substitutions in either binder or carbide has an effect on property values such as bend rupture strength, modulus, fracture energy absorbed, and hardness, as well as thermal conductivity and coefficient of expansion. A good bond usually occurs when the hard phase has a small amount of solubility in the binder phase during liquid phase sintering. Sessile drop experiments can be used to determine if the liquid binder phase wets the hard phase (ref 1). If one considers the shape of a molten drop of metal on a substrate, the balance of forces yields the expression

$$\gamma_{SV} = \gamma_{SL} + \gamma_{LV} \cos \theta \quad (1)$$

where  $\theta$  is the contact angle, and  $\gamma_{SV}$ ,  $\gamma_{SL}$ , and  $\gamma_{LV}$  are the surface energies (or surface tensions) of the solid-vapor, solid-liquid, and liquid-vapor interfaces, respectively. Wetting is said to occur if the contact angle is 90 degrees or less, and nonwetting occurs if it is 90 degrees or more where  $\cos \theta = \gamma_{SV} - \gamma_{SL} / \gamma_{LV}$ . Decreasing only the liquid surface tension ( $\gamma_{LV}$ ) of a given system by modification of the liquid phase reduces the contact angle if it is initially

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References are listed at the end of this report.

less than 90 degrees. This type of information is helpful to predict if a loss or a retention of the liquid phase will occur during sintering. Sweating, or a loss of the liquid phase from the compact, will occur in those systems exhibiting a contact angle larger than 90 degrees, while for contact angles less than this, the liquid will be retained. The bond strength or work of adhesion describing the bonding force between the solidified liquid and solid phases is given by the expression

$$W_a = \gamma_{SV} + \gamma_{LV} - \gamma_{SL} \quad (2a)$$

or

$$W_a = \gamma_{LV}(1 + \cos \theta) \quad (2b)$$

The data of Livey and Murray (ref 2) have revealed that as the contact angle decreases, the work of adhesion ( $W_a$ ) increases. Since copper does not react with carbides in general, it was used to find the relationships of these parameters for metallic carbides shown in Figure 1a. The heats of formation given in Table I for metallic carbides also have an effect on the wetting angle. The carbides with the higher negative heats of formation result in larger wetting angles than the carbides with lower heats of formation, e.g., molybdenum, carbon, or tungsten carbide as shown in Figure 1b.

TABLE I. HEATS OF FORMATION OF MONOCARBIDES\*  $-\Delta H_{298}$  kJ/mol

Titanium Carbide (TiC)	183.7
Zirconium Carbide (ZrC)	199.2
Hafnium Carbide (HfC)	209.2
Vanadium Carbide (VC)	126.4
Niobium Carbide (NbC)	142.3
Tantalum Carbide (TaC)	161.1
Chromium Carbide (Cr <sub>3</sub> C <sub>2</sub> )	87.9
Molybdenum Carbide (Mo <sub>2</sub> C)	17.6
Tungsten Carbide (WC)	35.1

\*References 3-5



Calculated bond strengths between the sessile drop of liquid binder metal, after being cooled and solidified, and the carbide substrate are rarely realized. This is because of the presence of microcracks or voids at the interface, or because angular particle shapes lead to stress concentrations. Kieffer (ref 6) has suggested that the higher strengths obtained for tungsten carbide with a cobalt binder are due to a blunting of imperfections in the carbide, and later reprecipitation occurring in a less damaging configuration. When a liquid is in contact with a solid for which there is appreciable solubility in the liquid, a diffusion gradient is set up across the interface and the interfacial energy can change as the composition of the liquid changes. The effect of compositional changes on interfacial energy is found in the titanium carbide-nickel system. When a nickel slug is placed on a titanium carbide disk and simultaneously heated at 1290°C, a eutectic liquid forms and the contact angle measures 30 degrees in vacuum at 1455°C. However, if a droplet of molten nickel is placed on the surface of the same carbide, it spreads rapidly and the contact angle measures only 4 degrees. The difference in contact angle is due to the lower interfacial energy at the titanium carbide-nickel interface compared to that of the eutectic liquid-carbide interface.

Another system which offers evidence that reductions in solid-liquid interfacial energy can improve wetting, and which is associated with a diffusion gradient across the interface, is the titanium carbide-molybdenum-nickel system. Molybdenum added to nickel was shown to decrease the alloy-carbide interfacial energy (ref 7). Molybdenum diffuses into the titanium carbide where the solubility of the molybdenum into the titanium is appreciable.

The microstructure resulting from a combination of carbide and binder phase has an effect on the properties obtained. A thin film of binder phase separating the carbide particles prevents a continuous fracture path through the brittle hard carbide phase. The thin film becomes less ductile and also increases the overall strength. Gurland and Bardzil (ref 8) studied the tungsten carbide-cobalt system and found that strength increases with decreasing binder thickness reaching a maximum at a mean free path of 0.5 micron between particles. Smith (ref 9) studied conditions leading to the dihedral angle ( $\phi$ ) formed by the junction of two adjacent grains at the liquid interface approaching equilibrium conditions. This angle is determined by interface energies such that

$$\cos \phi/2 = \gamma_{GB}/2\gamma_{SL} \quad (3)$$

where  $\gamma_{GB}$  is the grain boundary energy and  $\gamma_{SL}$  is the liquid-solid interface energy. The angle will be zero if  $\gamma_{SL} = \frac{1}{2}\gamma_{GB}$ , thus insuring that the continuous metal binder envelops the hard carbide grains.

A heavy metal theory proposed in the past (ref 10) tried to explain densification, grain growth, and phase distribution during liquid phase sintering. The process was thought to occur by solution of finer hard particles into the liquid and reprecipitation on large particles. Grain growth would be favored by a liquid in which solubility of the hard phase is large. This theory did not prove true in the tungsten carbide-copper system where there is little solubility of tungsten carbide in copper, but rapid carbide grain growth occurred. However, in the tungsten carbide-cobalt system where there is appreciable solution of the carbide in the cobalt liquid, little grain growth was evident. The

phenomenon described was rationalized by Parikh and Humenik (ref 11). They indicated if the liquid-solid interface energy is high as evidenced by a large contact angle using sessile drop testing, the hard particles are thought to coalesce under the action of surface tension forces, and grain growth occurs. This may be promoted further by a solution and reprecipitation process if that also occurs. When the liquid-solid interface energy is low, and where the contact angle is small, the hard phase is completely dispersed in the liquid. Therefore, little grain growth occurs by a solution and reprecipitation process even if there is appreciable solubility of the hard phase in the liquid.

## EXPERIMENTAL METHOD

### Materials

The heats of formation of monocarbides as shown in Table I were used to choose carbides from the upper, middle, and lower heat of formation values. Those chosen for this study were titanium carbide, zirconium carbide, vanadium carbide, and tungsten carbide. Since cobalt, as a liquid phase binder for carbides, has been investigated to a great extent in the past and is considered a strategic metal in times of a national emergency, nickel was chosen as the matrix binder to which increasing amounts of molybdenum were added up to 50 weight percent. The tungsten carbide and vanadium carbide powders were supplied by Hermann C. Starck, New York City. The molybdenum powder was supplied by General Electric Company, Cleveland, Ohio. The nickel, titanium carbide, and zirconium carbide were supplied by J. Matthey, Seabrook, New Hampshire. All powders were in the 1-4 micron range with oxygen contents of about 0.8 ppm for carbides and about 20 ppm for the metals.

## Procedures

Master blends of the nickel-molybdenum alloy binder were mixed on a weight percent basis and incorporated in binder amounts used, e.g., 10, 20, and 40 weight percent of the total carbide make up, as follows: A: 6.25 weight percent molybdenum, B: 12.5 weight percent molybdenum, C: 25 weight percent molybdenum, E: 50 weight percent molybdenum, the balance being nickel in each case. Originally a binder designated D (37.5 weight percent molybdenum, the balance being nickel) was included for study but was later dropped. The amount of carbide used in making up specimens was 90, 80, and 60 weight percent for each carbide chosen.

Mixing of the powders for the binder composition was done by dry milling for 1 hour. The correct proportions of carbide and binder were made up and dry milled together for 1 hour. The lubricant was made by mixing paraffin and hexane, a solvent, and heating it under an exhaust hood until all the paraffin dissolved. Each weighed powder batch was evaporated off, yielding 1.25 percent of lubricant.

Pressing of the powders was done in a double acting die press. After cold pressing at 50 Ksi (345 MPa), the powder compacts were placed on graphite plates. Dewaxing took place in a hydrogen atmosphere at 430°C and held at that temperature for 1 hour. Sintering was accomplished using a vacuum furnace held at a pressure of  $10^{-6}$  mm. The temperature was raised to 1200°C and maintained for 1 hour before being raised to the standard sintering temperature of 1450°C and held for 1 hour. The exceptions to the standard sintering temperature used were as follows: titanium carbide, all series with E binder, 1380°C was used; vanadium carbide, 80-20 series with binders A through E, 1350°C was used; while for the 60-40 series with binders A through N and the other two series using the N binder, 1270°C was used; tungsten carbide, 60-40 series with binders A through

E and with the other series using the N binder, 1290°C was used. The above changes were made because of obvious adverse reactions which occurred such as loss of specimen shape and fusion with the graphite substrate.

Hot isostatic pressing (HIP) of the previously sintered specimens was done on a contract basis by Gorham Industries, Gorham, Maine. The HIP pressure was at 15 Ksi (103 MPa) using argon, while the temperatures used were at 110°C below the previous sintering temperatures. Specimens were not coated prior to HIP so that any internal voids connected to the surface did not allow further densification to occur. Sizes of the specimens, which were of rectangular cross section, measured 4 mm by 5 mm wide, with a length of 4.45 cm (1.75 inches). The specimens, after removal from the sintering furnace, were measured for shrinkage, density, and hardness before testing. Three-point bend tests were conducted on all specimens after sintering and after HIP. The distance between end supports was 4.0 cm (1.576 inches). A differential transformer gage was used to measure deflection, and this, combined with the output of the load cell, allowed the plotting of load deflection curves. A floor model Instron Universal Testing Machine was used at a cross-head speed of 0.05 mm/min (0.002 in./min).

Fracture surfaces on selected specimens were examined using scanning electron microscope (SEM) fractography methods. Metallographic specimens were cut from cross sections of samples and polished and etched before photographing. Murakami's etch was most often used while a chromic oxide etch (0.5 gram chromic oxide in 100 ml hydrogen chloride) was used for the nickel-molybdenum alloy binders having eutectic microstructures.

## RESULTS

The microstructures resulting from additions of molybdenum to nickel in forming the liquid phase binder during sintering at 1450°C are shown in Figure 2a. Binder alloys N, A, B, and C are all single phase, while binder E has a eutectic structure. The E binder composition at 50 weight percent molybdenum-50 weight percent nickel falls to the right of the eutectic point at 47 weight percent molybdenum and to the left of the peritectic which forms at 62 weight percent molybdenum according to the nickel-molybdenum phase diagram (ref 12). The hardness values for the above binder alloys are shown in Figure 2b. The hardness values for the E binder average 955 Knoop hardness because of the features shown in Figures 2a and 3a. The crosslike formations are hard and measure 1128 Knoop compared to the background hardness of 697 Knoop. Cracks were observed emanating from the harder constitute. High molybdenum content in the crosslike formation is evident from x-ray analysis as shown in Figures 3b and 3c as compared to the background shown in Figure 3d. The same binder alloy sintered at 1380°C did not develop the crosslike formations and indicated an overall hardness of 860 Knoop. Microstructure and analysis of these features are shown in Figure 4 where a more uniform structure and distribution of elements is evident. Another feature of using molybdenum-nickel alloy binders is the reduction in weight loss due to evaporation brought about by sintering in vacuum shown in Figure 5. Pure nickel experienced a weight loss of 12.2 percent at 1450°C, while for alloy binders this loss was reduced anywhere from about one-half to one-fourth that for nickel.

The microstructures of titanium carbide sintered in vacuum at 1450°C are shown in Figure 6a. Porosity was present and only 84.5 percent of the theoretical density overall was attained. A refinement in microstructures is evident

using the alloy binders compared to the nickel binder. HIP of the previously sintered carbides reduced porosity markedly as shown in Figure 6b and greater than 90 percent of the theoretical density was attained. The microstructures of sintered tungsten carbide shown in Figure 7a indicate very little porosity present for those made with binders containing 6.25 and 12.5 weight percent molybdenum which attained about 97.0 percent of theoretical density; after HIP this was increased to just below 100 percent. HIP of tungsten carbide containing either higher concentrations of molybdenum in the binder or nickel alone did not reduce porosity as seen in Figure 7b and only 88 percent of the theoretical density was achieved. A refinement in microstructure was brought about by using the alloy binder compared to using only nickel. The binder which forms a eutectic microstructure as shown previously in Figure 2a, however, does not provide a refinement in carbide structure or a reduction in porosity in either the sintered or HIP conditions for tungsten carbide shown in Figure 7. The microstructures of vanadium and zirconium carbides were not closely documented. Vanadium carbide, with a theoretical density of 89.7 percent, had better mechanical properties than zirconium carbide, with a theoretical density of 82.0 percent. However, there was little improvement in densities or mechanical properties in either one after HIP. Fracture surfaces from bars broken during bend testing before and after HIP are shown in Figure 8. Densification after HIP is especially noticeable in Figure 8b for titanium carbide, where in the sintered condition, porosity and nonwetting were evident prior to HIP. Fracture surfaces from tungsten carbide specimens parallel behavior shown in the micrographs of Figures 7a and b.

The mechanical properties obtained from bend testing were flexural modulus, fracture energy, bend rupture strength, and hardness. These parameters are shown in Figures 9 through 14 for the 90-10, 80-20, and 60-40 series of carbides

tested. The data for the 80-20 series, shown in Figures 11 and 12, appear more meaningful compared to the others. A peaking of modulus values and, to a lesser extent, fracture energy for titanium carbide and tungsten carbide fabricated with binder alloys A and B (6.25 and 12.5 weight percent molybdenum) is evident as shown in Figures 11a and b. The bend rupture strength and hardness values follow this same trend shown in Figures 12a and b.

## DISCUSSION

A decrease in the liquid-solid interfacial energy is a method of improving wetting between the liquid phase and carbides even in systems where wetting is considered marginal at best and large contact angles are measured. Improvement in wetting can occur where there is preferential absorption of atoms at an interface lowering interfacial energy. Whalen and Humenik (ref 1) indicated this occurs when nickel from the nickel-copper alloy binder is absorbed at the titanium carbide interface. Also, when there is a diffusion gradient established across solid-liquid interfaces, contact angles are reduced, improving wetting. The wetting of graphite by an iron-carbon alloy also illustrates this approach. When the interfacial energy is low and the carbon diffuses across the interface, good wetting occurs, the extent of which depends on the amount of carbon that is in solution in the iron. Both the interfacial energy and the contact angles increase rapidly when the equilibrium content of carbon in iron is reached and poor wetting results.

Considering the nickel-molybdenum alloy system in wetting of titanium carbide, molybdenum added to nickel was found to decrease the interfacial energy (ref 7). The molybdenum was found to diffuse into the carbide quite readily because of complete liquid and solid solubility of molybdenum in titanium.



Molybdenum and tungsten also form a continuous series of solid solutions and indicate the same for the liquid state. Molybdenum has only limited solid solubility in zirconium and vanadium. The response of the nickel-molybdenum binder used in this study appears to depend on the diffusion of molybdenum into the carbide. The binder alloys were made up by increasing the amounts of molybdenum added to nickel until a nearly eutectic composition was reached at 50 weight percent molybdenum-50 weight percent nickel. When this alloy was used, the molybdenum was effectively tied up and diffusion across the interface was reduced; thus wetting of the carbides was lessened. Strength and ductility attained by liquid phase sintering of heavy metals have been shown to be inversely related to contiguity between the hard phases indicated by German and Bourguignon (ref 13). Reduction in wettability is inferred by an increasing number of contacts between the hard, nearly insoluble particles which subsequently reduce strength values. Better wetting of the solid by the liquid results in lower contiguity and lower dihedral angles with higher strength values attained. Using bend rupture strength in this study as a measure of wettability by the liquid phase, we see that strength values generally decreased the closer the binder alloy composition approached the eutectic composition. This was found true for 90-10, 80-20, and 60-40 series as shown in Figures 10, 12, and 14. The wetting of carbides by a liquid metal as measured by the contact angle was shown to be inversely proportional to their heats of formation (ref 14). The more stable the carbide was, the less the tendency for wetting. The above study by Ramquist indicated this as true using nickel, cobalt, and iron as the liquid binder in the wetting of carbides with heats of formation from -209 kJ/mol for hafnium carbide to -17.6 kJ/mol for molybdenum carbide. Neither contact nor dihedral angles were measured in the wetting of carbides by

the molybdenum-nickel binder used in the present study, but results from mechanical testing infer that higher test values are the result of improved wetting.

The vacuum furnace used in the present study was not designed for sintering per se. Mesh tungsten heaters allowed elevated temperatures to be reached. However, a good vacuum had to be maintained for the interlock system to remain open and allow current flow to the heaters. As a result, weight losses due to evaporation are rather large, especially for nickel, as shown in Figure 5. Undue evaporation losses likely resulted in poorer properties being attained than could otherwise be realized if a lesser vacuum (e.g.,  $10^{-3}$  mm) were used during liquid phase sintering.

#### SUMMARY

1. The effectiveness of the nickel-molybdenum alloy binder in promoting small carbide grain size and improved mechanical properties was greatest at lower concentrations of molybdenum in nickel, e.g., 6.25 and 12.5 weight percent.

2. Wetting of some carbides was improved by using the nickel-molybdenum alloy liquid phase binder which established a diffusion gradient across the liquid-solid interface, whereby molybdenum was free to diffuse into the carbides, thus lowering interfacial energies.

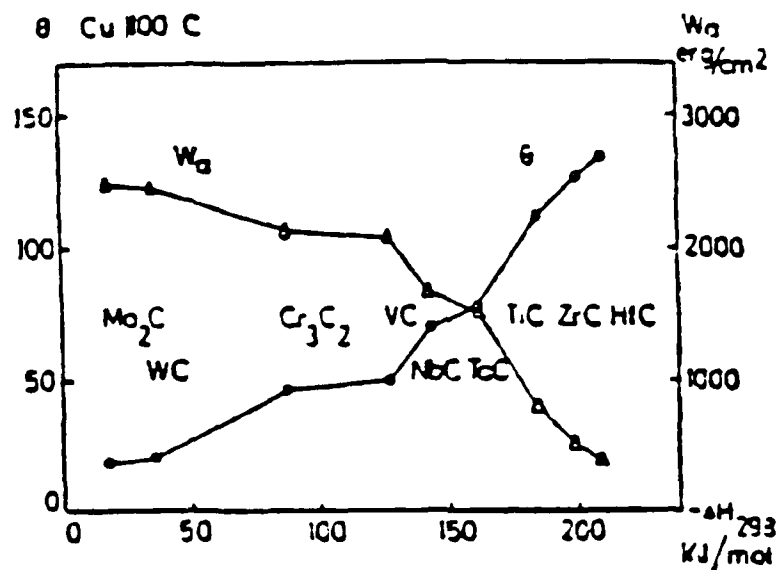
3. When the liquid phase alloy approached the eutectic composition, the diffusion of molybdenum into the carbides was halted, thus raising interfacial energies and reducing wetting of the carbides. Porosity remained, along with propensity towards crack formations, seen in the binder when melting occurred at 1450°C.

4. Evaporation losses in the liquid phase binder, due to melting in a vacuum at 1450°C, were reduced by one-fourth to one-half using the nickel-molybdenum alloy binder as compared to using unalloyed nickel.

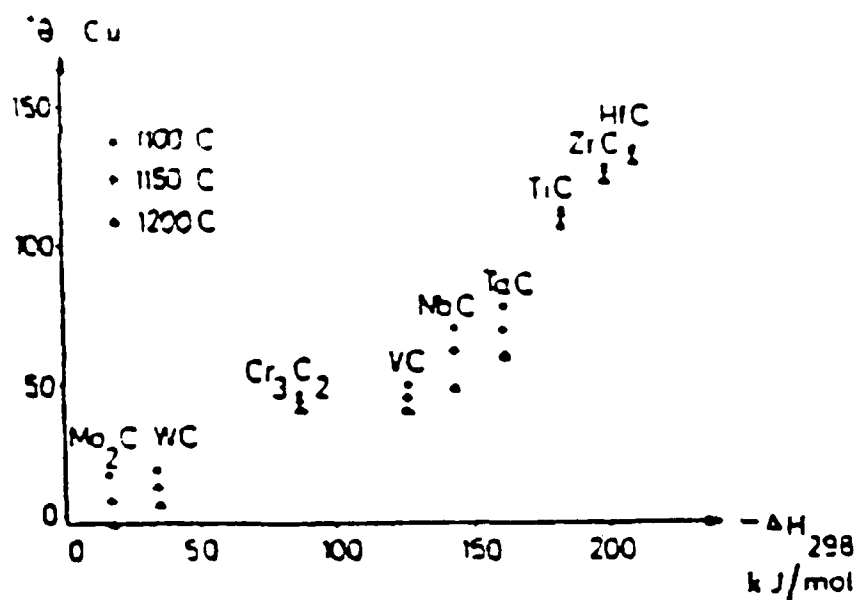
5. The effect of HIP densification at 15 Ksi and at 110°C below the sintering temperatures could not be evaluated equally for all the carbides because specimens were not encapsulated or coated prior to HIP.

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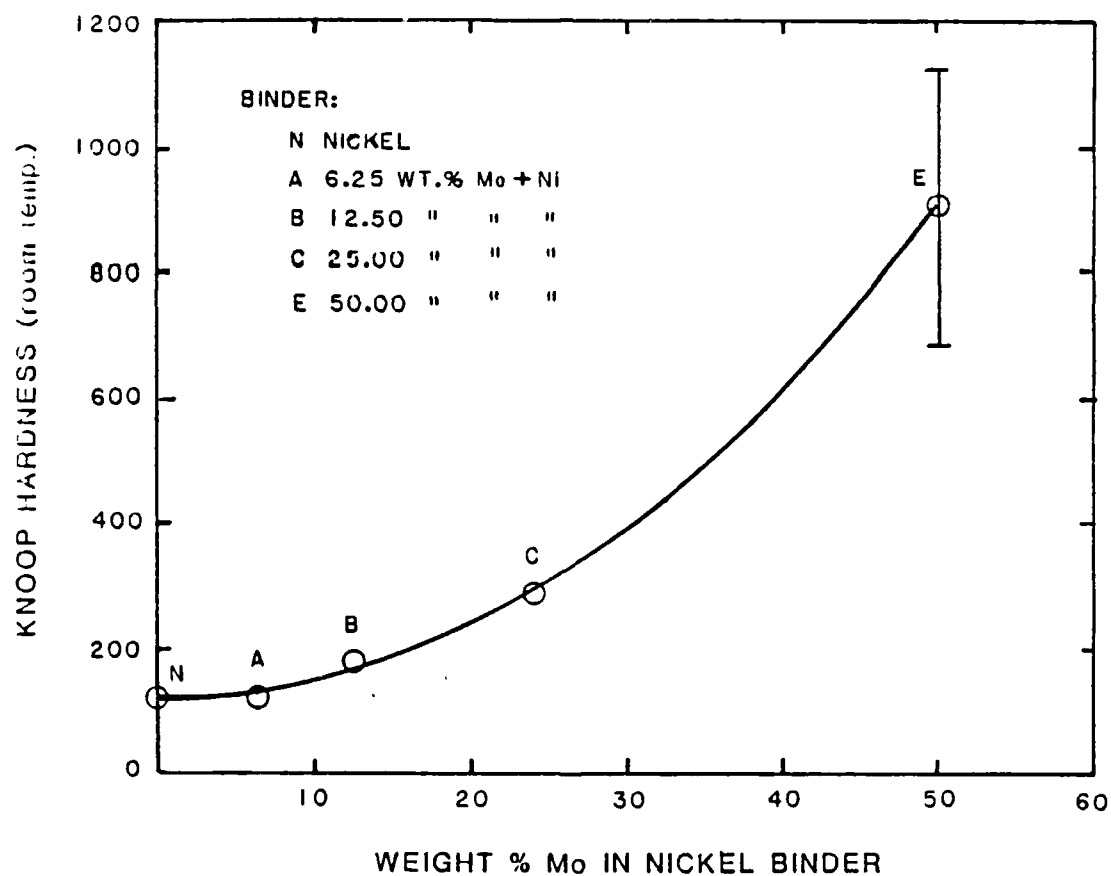
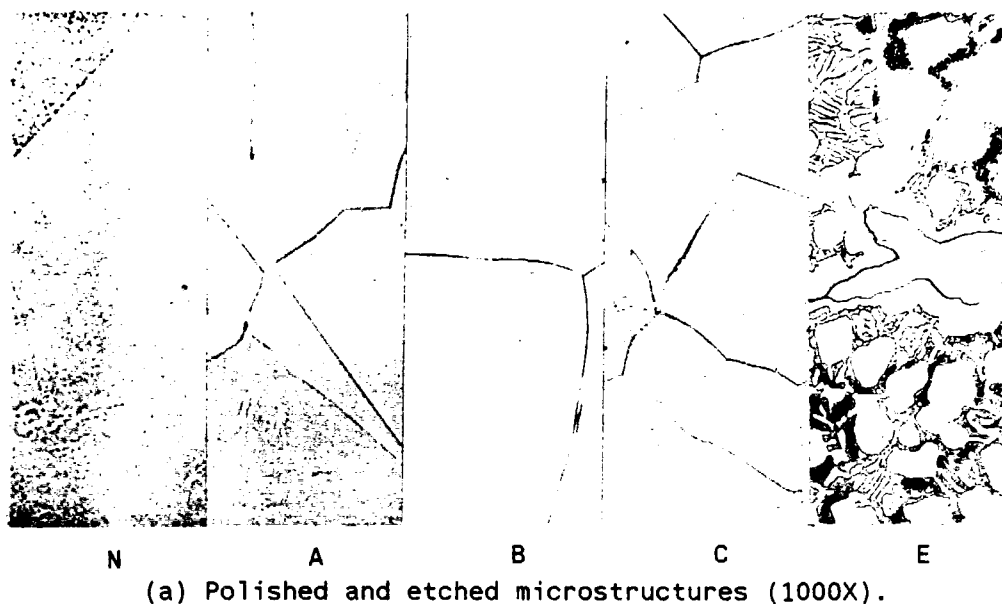


(a)



(b)

Figure 1. (a) The relationships between wetting angle, work of adhesion, and heat of formation for metallic carbides wetted in vacuum by copper at 1100°C. (b) The relationship between heats of formation and wetting angles for metallic carbides wetted by copper at 1100°C, 1150°C, and 1200°C (ref 14).

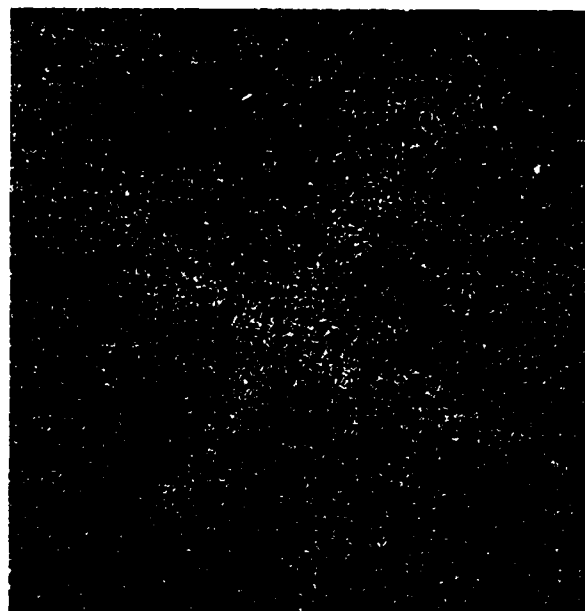


(b) Knoop hardness as a function of weight percent molybdenum added to nickel.

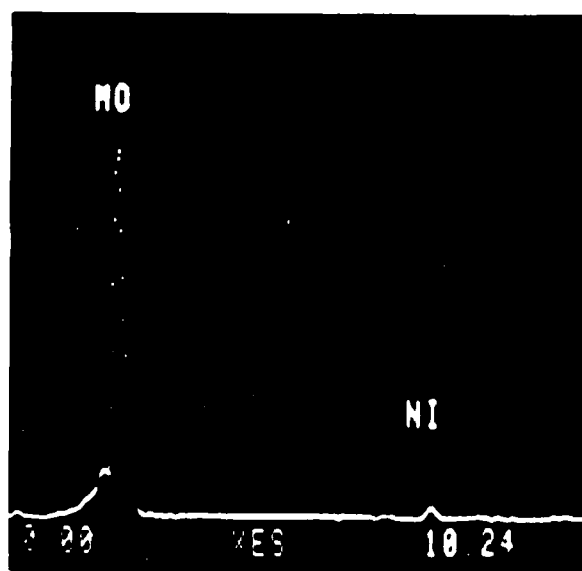
Figure 2. Microstructures and corresponding hardness for a nickel alloy binder with increasing weight percent molybdenum in the binder melted at 1450°C in vacuum.



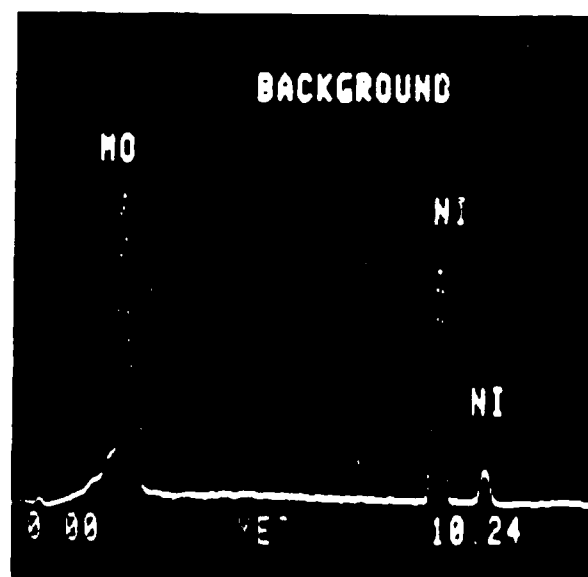
(a)



(b)

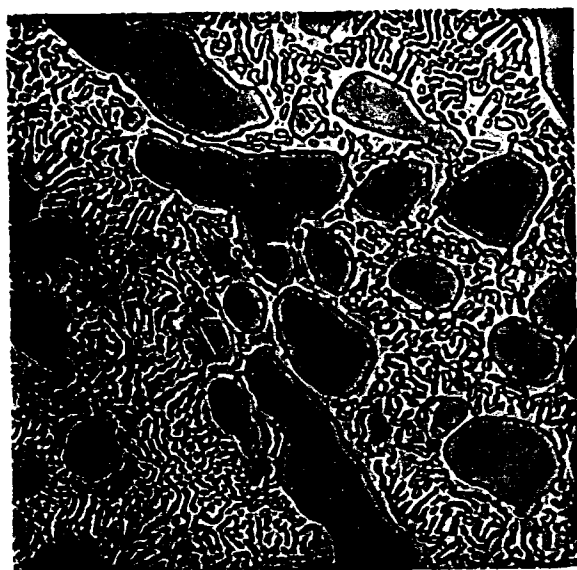


(c)

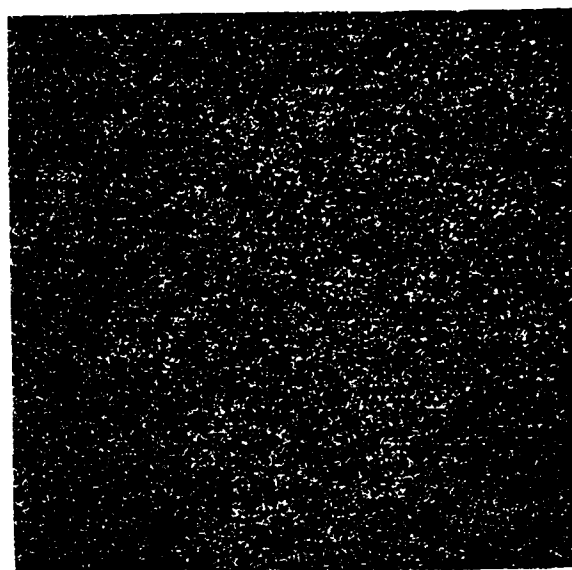


(d)

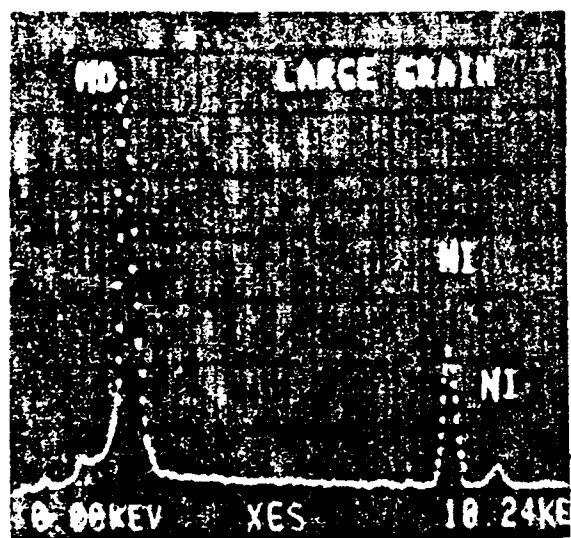
Figure 3. Microstructure and analysis of the nickel alloy binder designated E (eutectic) containing 50 weight percent molybdenum-50 weight percent nickel and melted at 1450°C in vacuum.  
 (a) Polished and etched (1000X).  
 (b) X-ray dot mapping indicating high molybdenum content.  
 (c) and (d) Energy dispersive microanalyses using x-rays for cross-like formation and background.



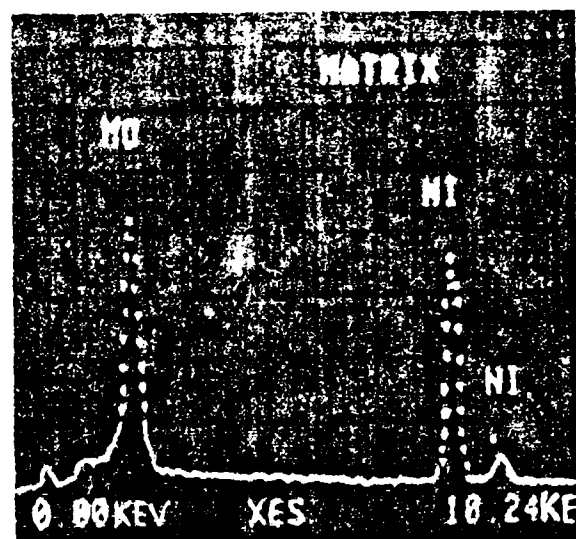
(a)



(b)



(c)



(d)

Figure 4. Microstructure and analysis of the nickel alloy binder designated E (eutectic) containing 50 weight percent molybdenum-50 weight percent nickel and melted at 1380°C in vacuum.

(a) Polished and etched (1000X).

(b) X-ray dot mapping of the area shown.

(c) and (d) Energy dispersive microanalyses using x-rays for the primary dendrites and background, respectively.



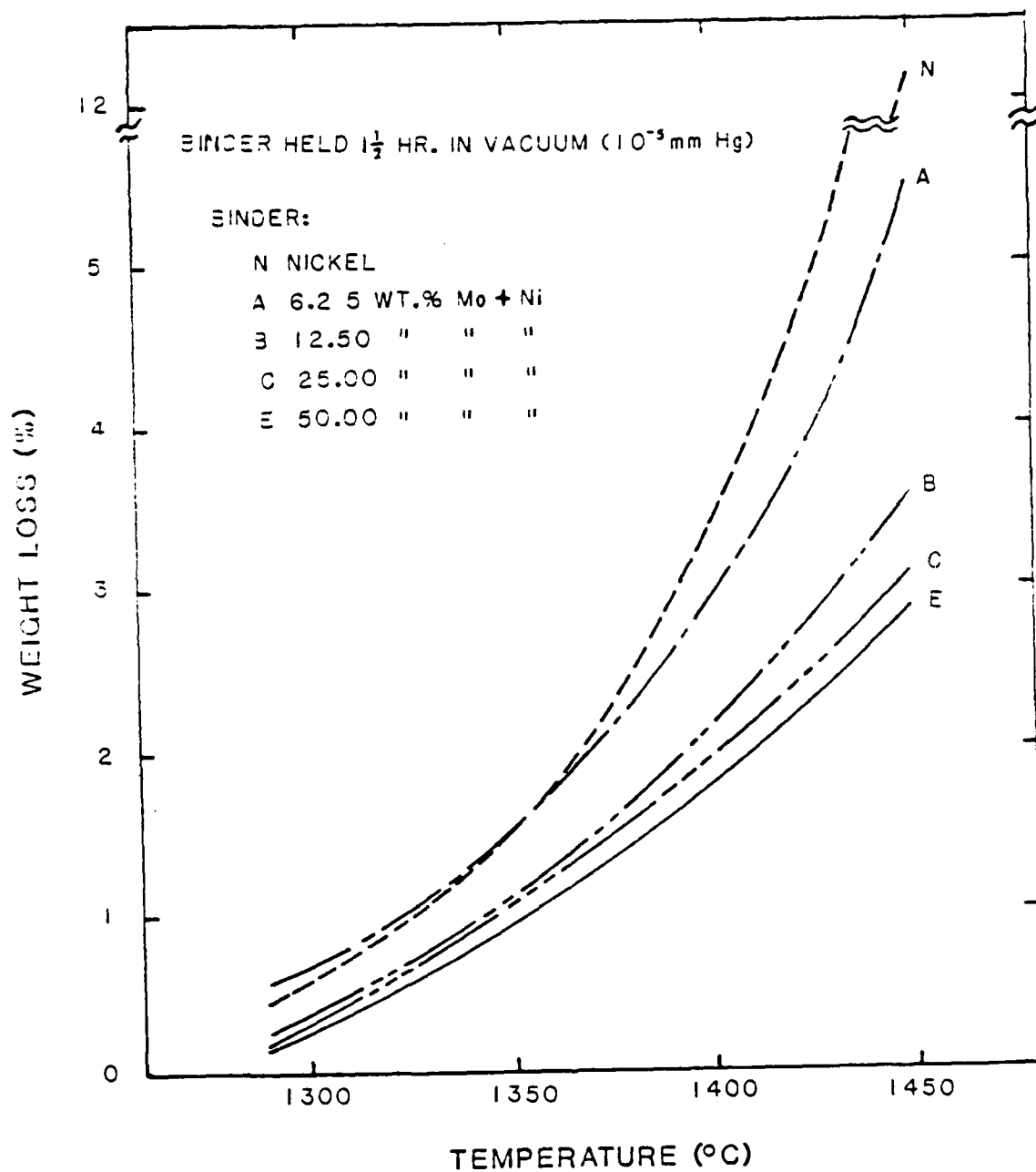
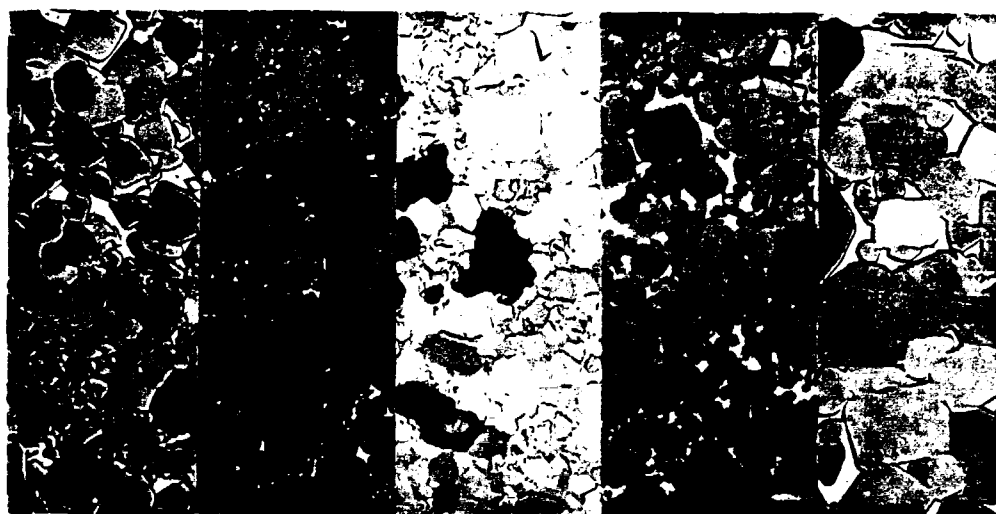
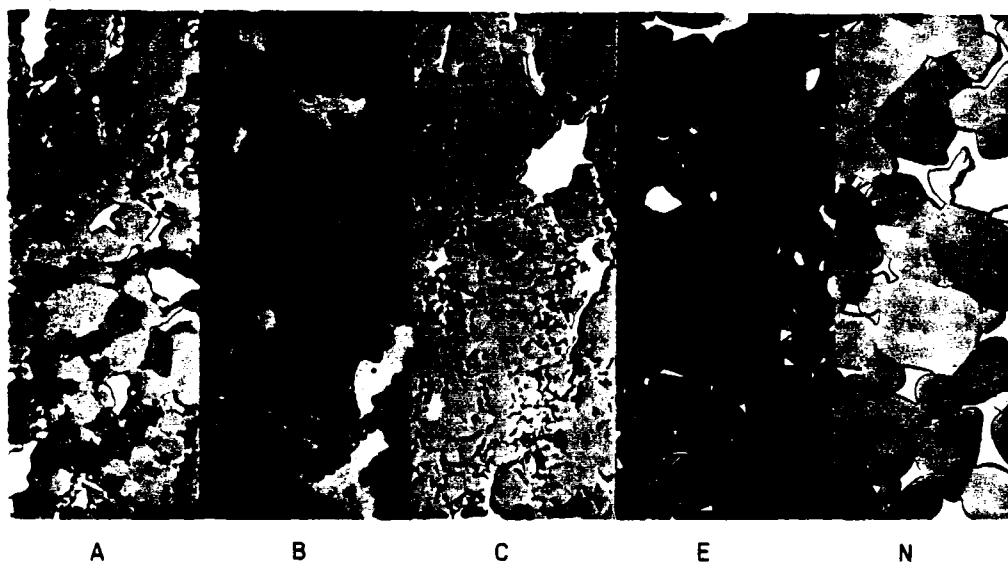


Figure 5. Weight loss in nickel-molybdenum alloy binders by evaporation after being held for  $1\frac{1}{2}$  hours in vacuum at temperatures from 1290°C to 1450°C.



(a) Titanium carbide microstructures after sintering (1000X).

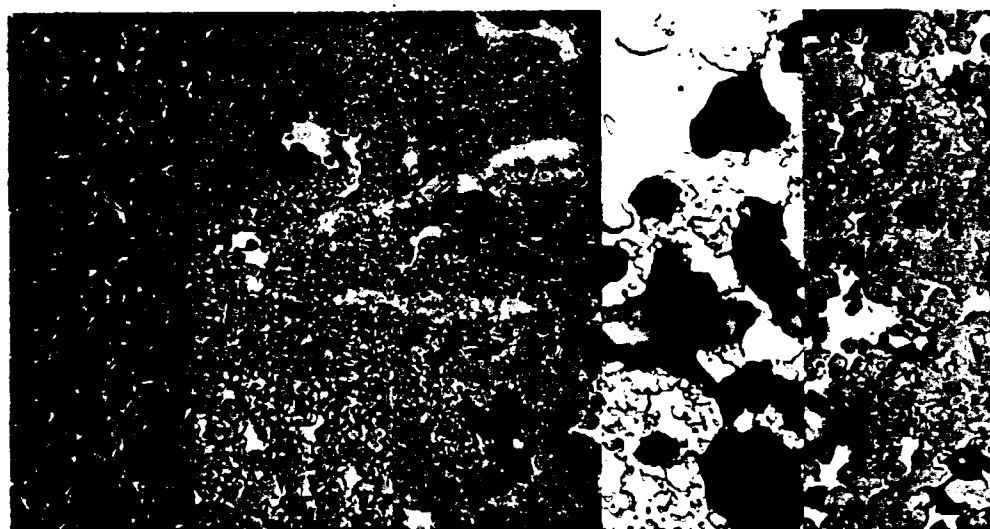


(b) Titanium carbide microstructures after sintering and HIP (1000X).

- Figure 6. (a) Titanium carbide (80 weight percent) sintered in vacuum at 1450°C for 1 hour with binders containing nickel and with increasing weight percents of molybdenum in the binder phase as follows: A: 6.25%, B: 12.50%, C: 25.00%, E: 50.00%, N: nickel only - 1290°C.
- (b) Reduction of voids after HIP of previously sintered titanium carbide (80 weight percent) with binders that contain nickel and molybdenum as above. HIP densified at 15 Ksi (103 MPa) and at 110°C below the previous sintering temperatures.

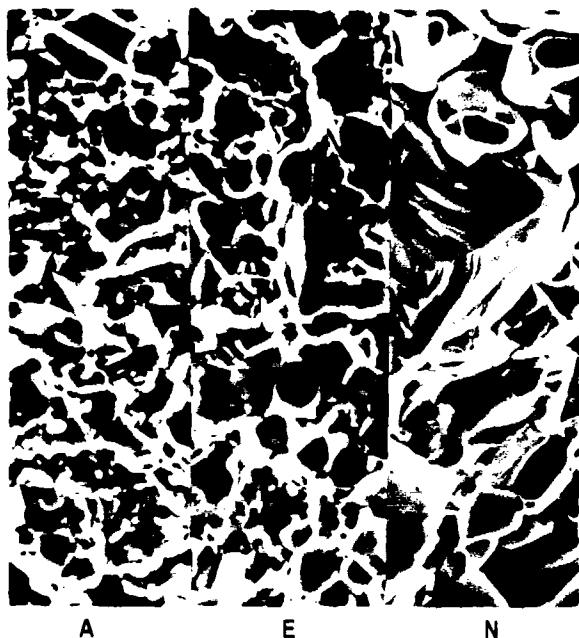


A B C E N  
(a) Tungsten carbide microstructures after sintering (1000X).

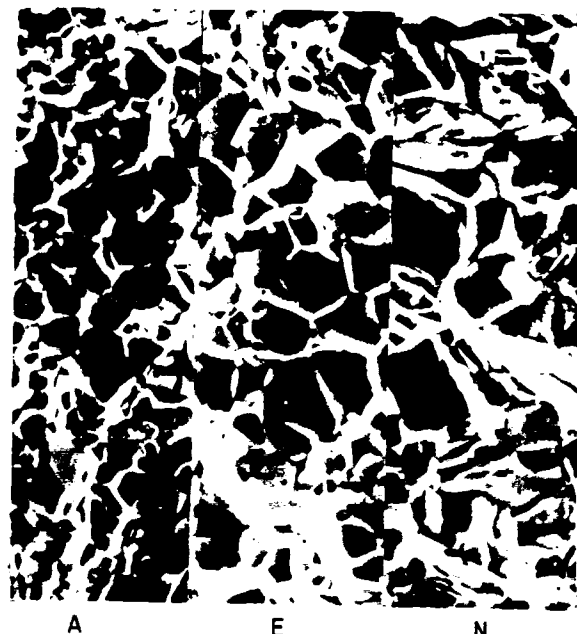


A B C E N  
(b) Tungsten carbide microstructures after sintering and HIP (1000X).

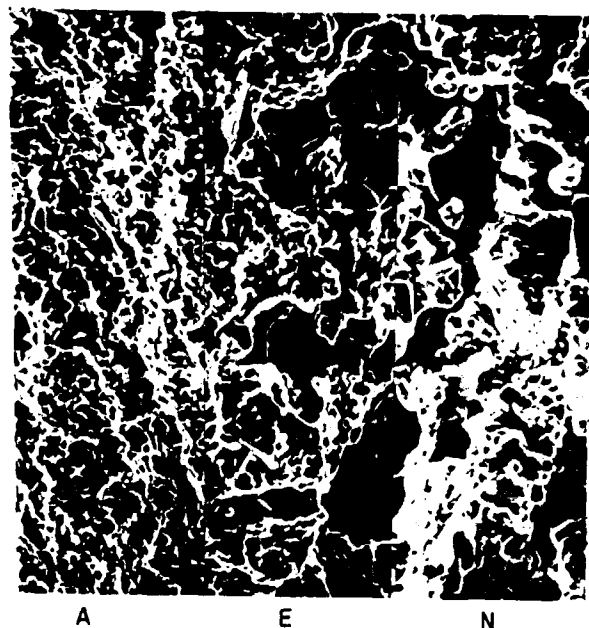
- Figure 7. (a) Tungsten carbide (80 weight percent) sintered in vacuum at 1450°C for 1 hour with binders containing nickel and with increasing weight percents of molybdenum in the binder phase as follows: A: 6.25%, B: 12.50%, C: 25.00%, E: 50.00%, N: nickel only - 1290°C.
- (b) Reduction of voids after HIP of previously sintered tungsten carbide (80 weight percent) with binders that contain nickel and molybdenum as above. HIP densified at 15 Ksi (103 MPa) and at 110°C below the previous sintering temperatures.



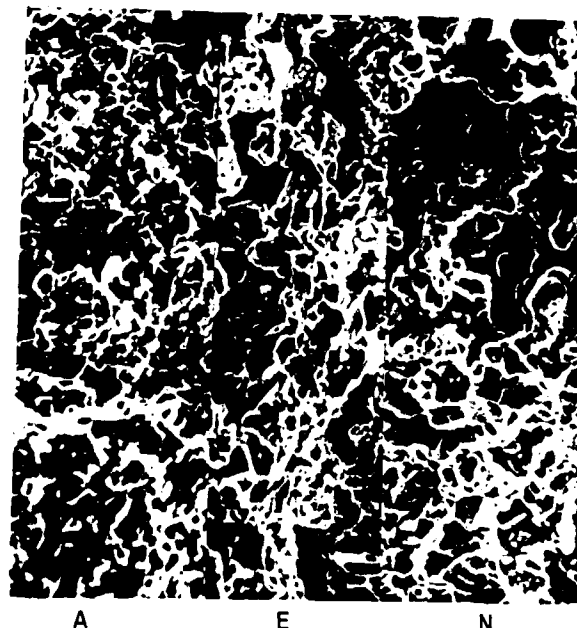
(a) Titanium carbide - sintered



(b) Titanium carbide - sintered and HIP

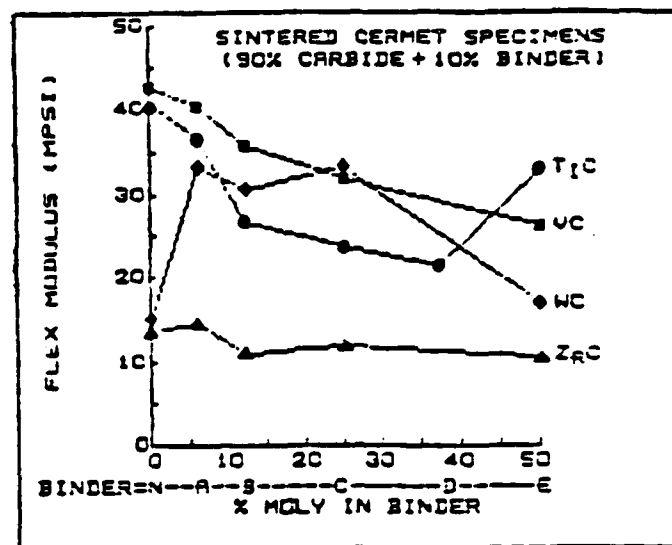


(c) Tungsten carbide - sintered

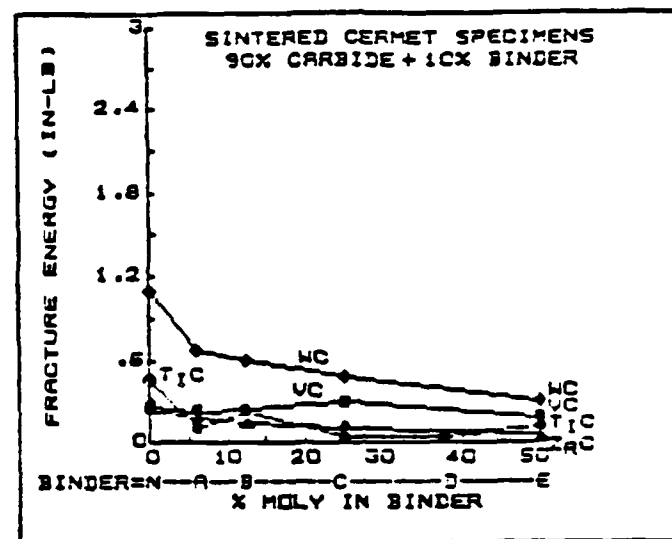


(d) Tungsten carbide - sintered and HIP

Figure 8. Bend fracture surfaces of carbides liquid phase sintered at 1450°C using binders containing nickel and with increasing weight percent of molybdenum as follows: A: 6.25%, E: 50.00%, N: 100% nickel.  
 (a) and (b) 80 titanium carbide-20 binder.  
 (c) and (d) 80 tungsten carbide-20 binder. (SEM 1500X)

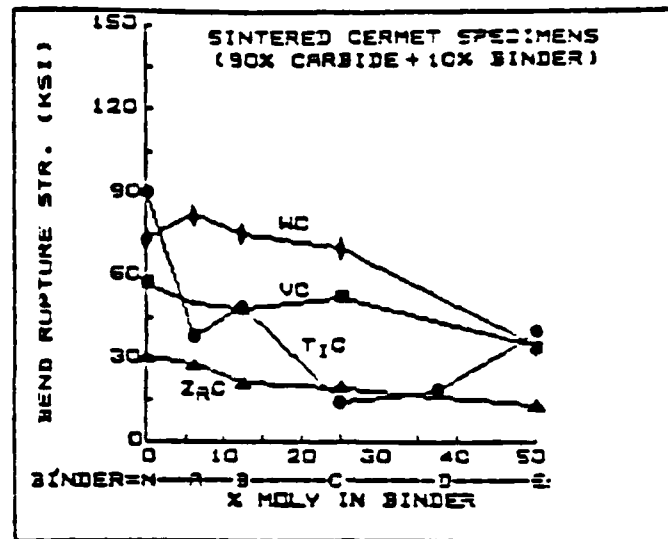


(a) Flexural modulus.

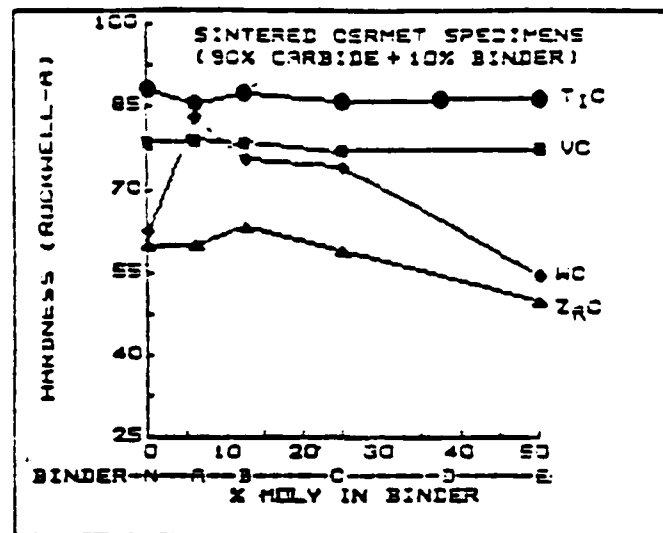


(b) Fracture energy.

Figure 9. Properties of the 90-10 series of carbides as a function of weight percent molybdenum alloyed with nickel in the binder.

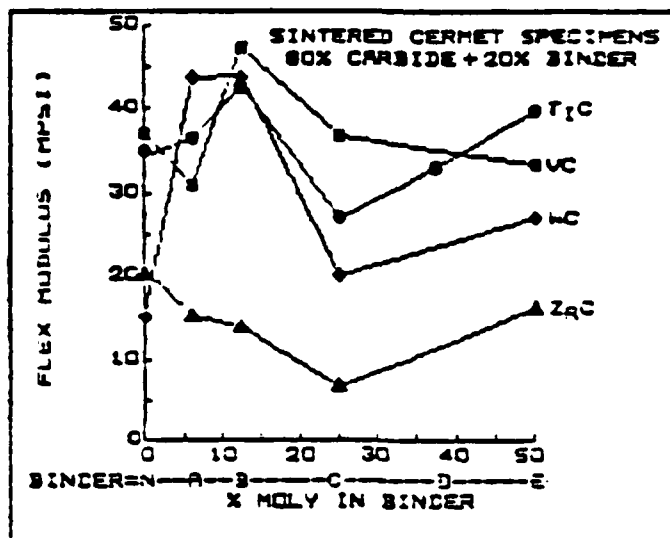


(a) Bend rupture strength.

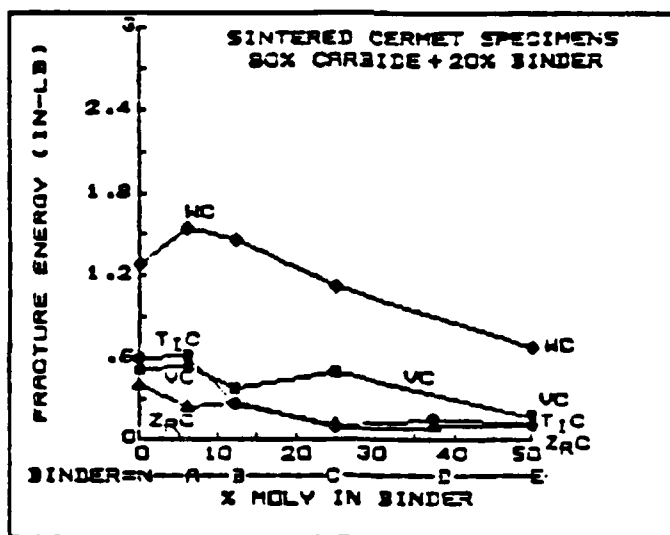


(b) Hardness.

Figure 10. Properties of the 90-10 series of carbides as a function of weight percent molybdenum alloyed with nickel in the binder.

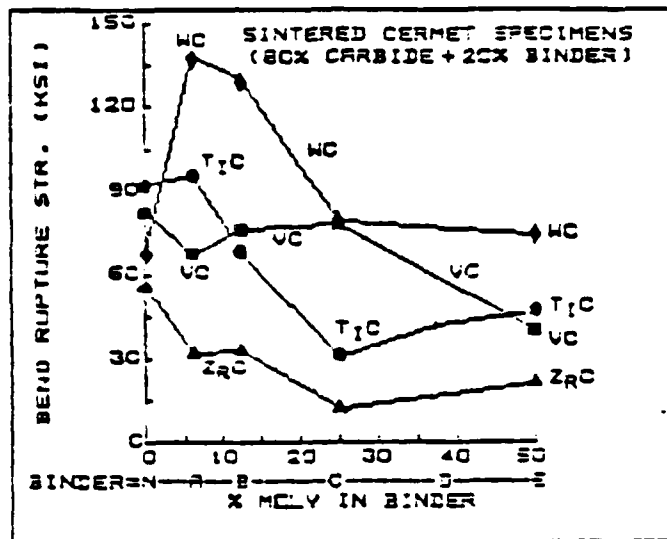


(a) Flexural modulus.

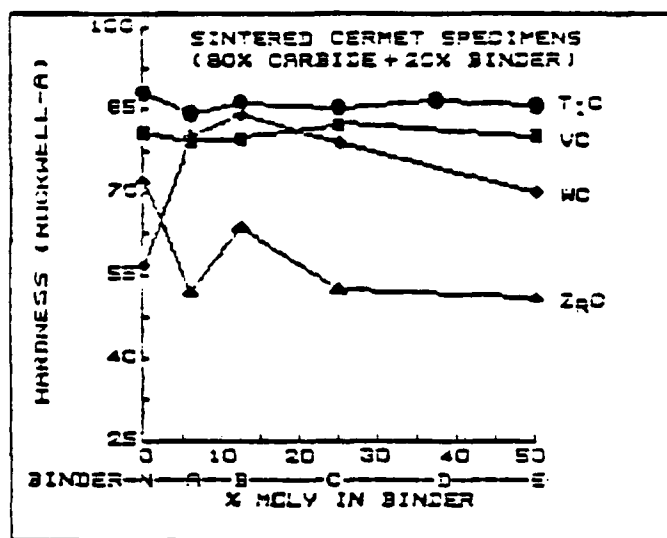


(b) Fracture energy.

Figure 11. Properties of the 80-20 series of carbides as a function of weight percent molybdenum alloyed with nickel in the binder.



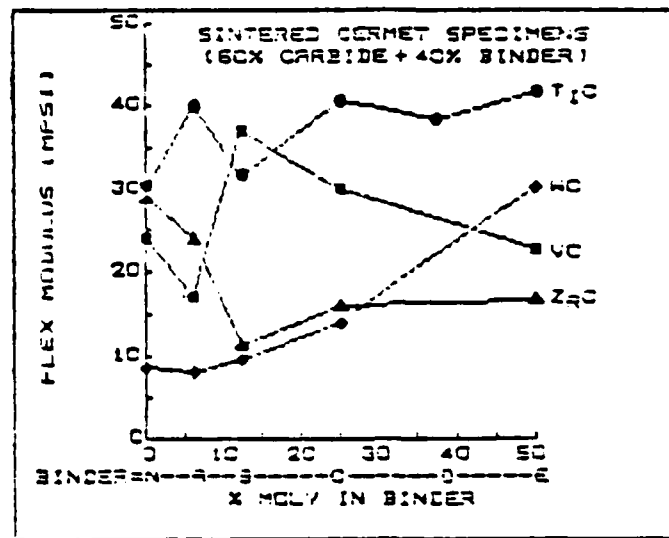
(a) Bend rupture strength.



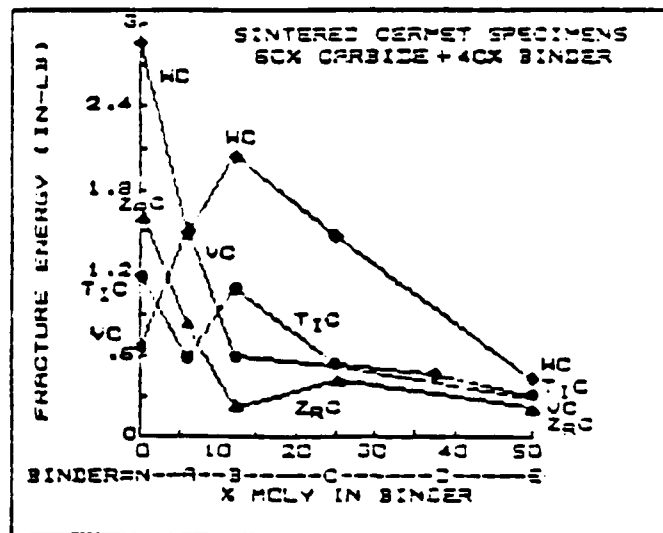
(b) Hardness.

Figure 12. Properties of the 80-20 series of carbides as a function of weight percent molybdenum alloyed with nickel in the binder.



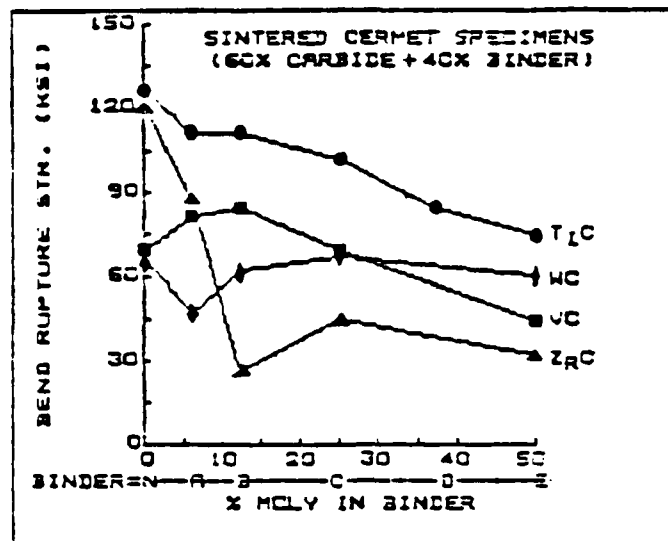


(a) Flexural modulus.

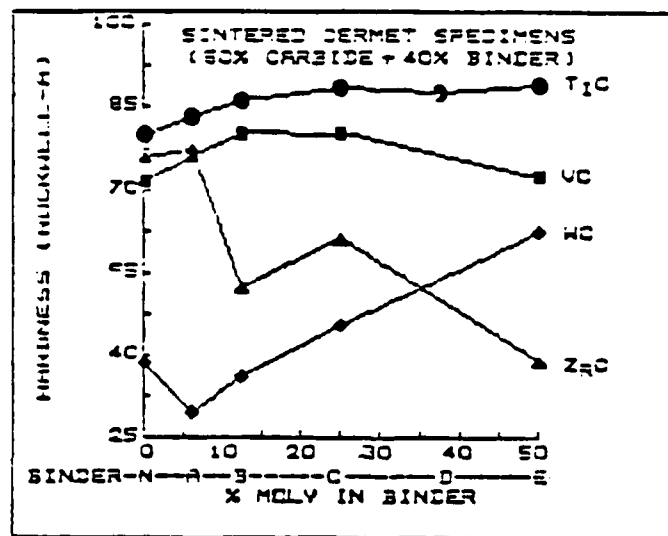


(b) Fracture energy.

Figure 13. Properties of the 60-40 series of carbides as a function of weight percent molybdenum alloyed with nickel in the binder.



(a) Bend rupture strength.



(b) Hardness.

Figure 14. Properties of the 60-40 series of carbides as a function of weight percent molybdenum alloyed with nickel in the binder.

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